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14. ABSTRACT

A study of fatigue damage in lamellar polycrystals of γ-TiAl based intermetallic alloys have been undertaken. Using electropolished fatigue specimens, it has been found that most of fatigue failures occur due to cracks initiating internally. Surface damage prior to failure has been very minimal. The specimens showed very little change in cyclic stress-strain hysteresis prior to failure. This mechanism of fatigue failure is atypical when compared to most conventional materials, and requires a very careful study both from a fundamental and engineering perspective. A theoretical model was develped to related the microcracks foming in fatigue, to the macroscopic ductility of γ-TiAl specimens

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FATIGUE OF LAMELLAR POLYCRYSTALS IN GAMMA TITANIUM ALUMINIDES

(AFOSR GRANT: F49620-00-1-0251)

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ABSTRACT

A study of fatigue damage in lamellar polycrystals of γ-TiAl based intermetallic alloys have been undertaken. Using electropolished fatigue specimens, it has been found that most of fatigue failures occur due to cracks initiating internally. Surface damage prior to failure has been very minimal. The specimens showed very little change in cyclic stress-strain hysteresis prior to failure. This mechanism of fatigue failure is atypical when compared to most conventional materials, and requires a very careful study both from a fundamental and engineeringpersp ective. A theoretical model was developed to relate the microcracks forming in fatigue, to the macroscopic ductility of γ-TiAl specimens

I. RESEARCH OBJECTIVE

The research involves a study of fatigue life behavior as well as the life partitioning between crack nucleation and growth stages of two fully lamellar microstructures of a γ-TiAl based intermetallic. Two microstructures that differed in the lamellarspacing, but have similar grain sizes were employed. The additional objective was to theoretically model the effect of microcracks created in fatigue, on the macroscopic ductility of the material.

II. MATERIAL

The as-received material was the TiAl alloy: Ti-46.5Al-1.5Cr-0.5Mn-3Nb-0.2W-0.2Hf-0.2Zr-0.2B-0.2C-0.2O_x (at.%). The processing history of the plate involved induction skull meltingto produce ingot and then hot isostatic pressing. A cylindrical billet was cut out from the ingot to and isothermally forged at 1150°C (with 91% reduction) to form pancake-shaped plate. S-N fatigue specimens were machined from the plate and were heat-treated to produce the two types of microstructural conditions.

III. EXPERIMENTAL

The heat treatment was done in a resistance heated tube furnace in an argon atmosphere to prevent surface oxidation. The specimens were also wrapped in several layers of titanium foil during heat treatment. Heat treatment schedules to produce the two different lamellar spacings (designated as fine-lamellarand coarse-lamellarstructures) were:

Fine-lamellar:Annealed at 1360°C/1 hr., furnace cooled to 1250°C at an initial cooling rate of about 60°C/minute, and then immediately air cooled to room temperature.
Coarse-lamellar:Annealed at 1360°C/1 hr., furnace cooled to 1250°C at an initial cooling rate of about 3°C/min and then air cooled to room temperature.
After heat treatment, surfaces of the S-N fatigue specimens were mechanically polished down to 600 grit to remove any contamination during heat treatment. The specimens were then electropolished to remove the layers affected by mechanical polishing and also to produce a microscopically smooth surface suitable for replication study during the fatigue experiment.

The S-N fatigue tests were conducted in a MTS-810 fatigue test system equipped with TESTSTAR II digital controller. The tests were performed in tension-tension at the stress ratio of 0.1 at room temperature and the lab air environment. A test frequency of 35 Hz was used except for recording of the hysteresis when the frequency was reduced to 5 Hz. Test was put on hold at the mean load at regular intervals to record replica of the surfaces.

RESULTS

IV. 1 Fatigue Data and Behavior

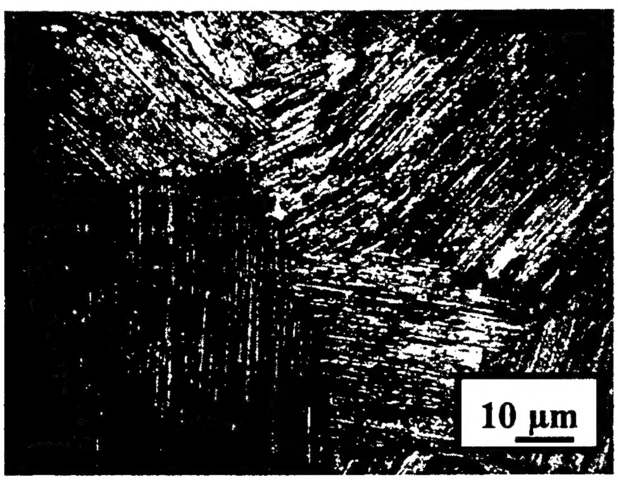
Optical micrographs of the resulting fine-lamellarand the coarse-lamellarmicrostructures are presented in Figs. 1(a) and (b) respectively. It is clear from Fig. 1 that the above heat treatment steps produced a considerable variation in the lamellar width of the two microstructures.

The fatigue life behavior of the fine-lamellarmicrostructure is shown in Fig. 2. The fatigue life ranged from about 1000 cycles at the stress level (σ_{max}) of 450 MPa to about $4x10^5$ cycles at the σ_{max} level of 425 MPa. The life to failure followed a straight-line profile (Fig. 2), at least in the life range of 1000 cycles to about $4x10^5$ cycles.

Surfaces of the S-N fatigue specimens were replicated at regular intervals. The aim was to capture the formation of slip bands as well as the first crack. However, slip bands could not

be resolved even at cycles close to the failure cycle of the specimen. It is possible that the fine lamellar nature of the microstructure masked the slip bands. For example, Fig. 3 shows surface replica of one of the surfaces of the specimen cycled at a σ_{max} level of 450 MPa. The replica was recorded after 5049 cycles while the specimen failed in 8274 cycles. However, there is no slip bands can be seen in the replica (Fig. 3). Also, a through examination of replicas from the two surfaces could not reveal any surface cracks even at 5049 cycles. In order to determine if the crack nucleated in the subsurface of the specimen or at a corner, the fracture surface was examinedin a scanning electron microscope (SEM). The fractographs of the specimen cycled at σ_{max} of 450 MPa are presented in Fig. 4 showing the crack nucleation region (Fig. 4(a)) and a region away from the crack origin (Fig. 4(b)). It is clear that crack started near a corner (Fig. 4(a)). Also, while crack propagation near the origin (Fig. 4(a)) seems to be intergranular, away from the origin (Fig. 4(a)) crack seems to have propagated by cleavageacross lamellacolonies.

(a)



(b)

Fig. 1: Microstructures generated in the γ-TiAl based material; (a) fine-lamellarstructure and (b) coarse-lamellarstructure

The stress vs. strain data was also recorded at regular cyclic intervals. It is of interest to examine if nucleation of crack caused an increase in the size of hysteresis thus enabling the detection of life for crack nucleation, irrespective of surface or subsurface nucleation. The hysteresis data from the specimen cycled at a σ_{max} of 460 MPa, that failed in 1087 cycles, are shown in Fig 5. Figures 5(a) and (b) show the hysteresis plots at 520 and 1060 cycles respectively. It is clear that there is very little hysteresis in these plots. Further, on comparison, not much difference can be seen between the plot at 520 cycles and the one at 1060 cycles. TiAl based intermetallics are known to have a very small damage tolerance. In other words, a crack may propagate in only a few cycles after nucleation, especially in low cycle fatigue regime as is the case in Fig. 5. It may, therefore, be difficult to detect a change in hysteresis when fatigue lives are in the range of 1000 cycles or less.

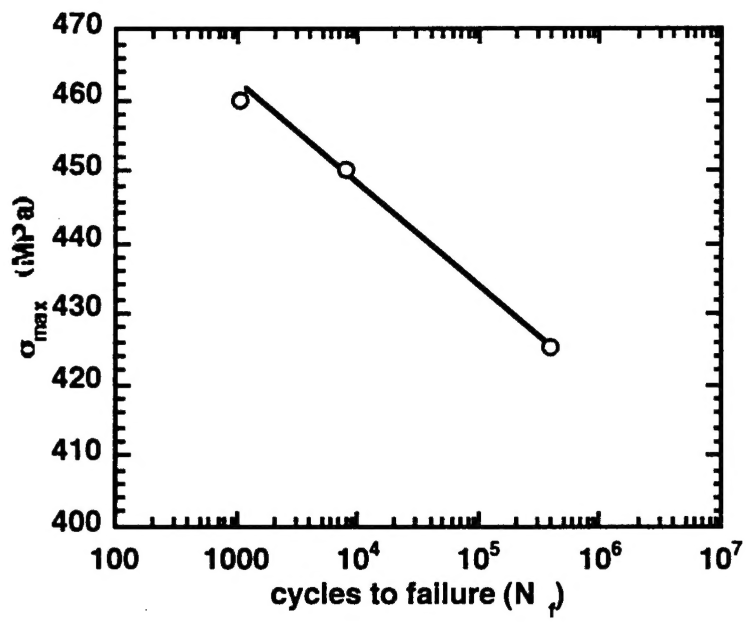


Fig. 2: S-N fatigue data from the fine-lamellar microstructure

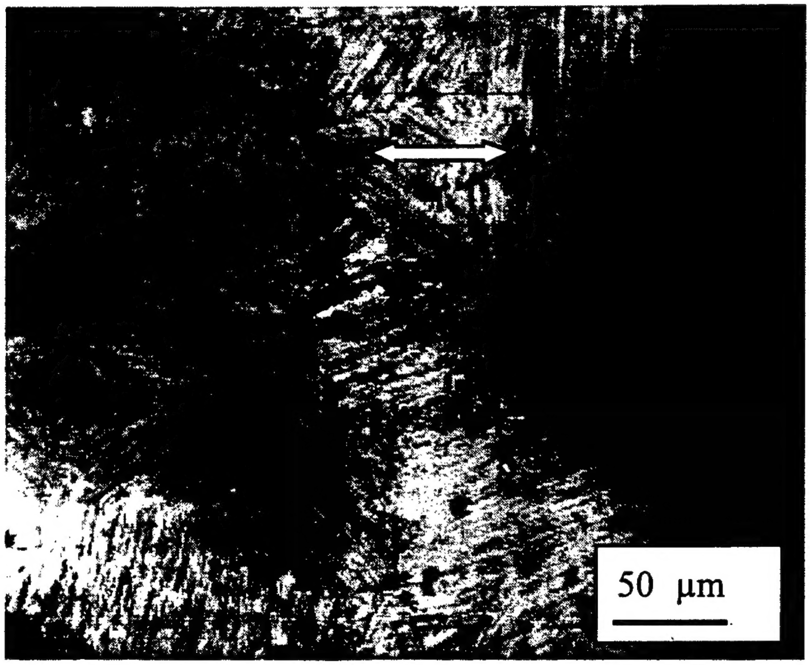
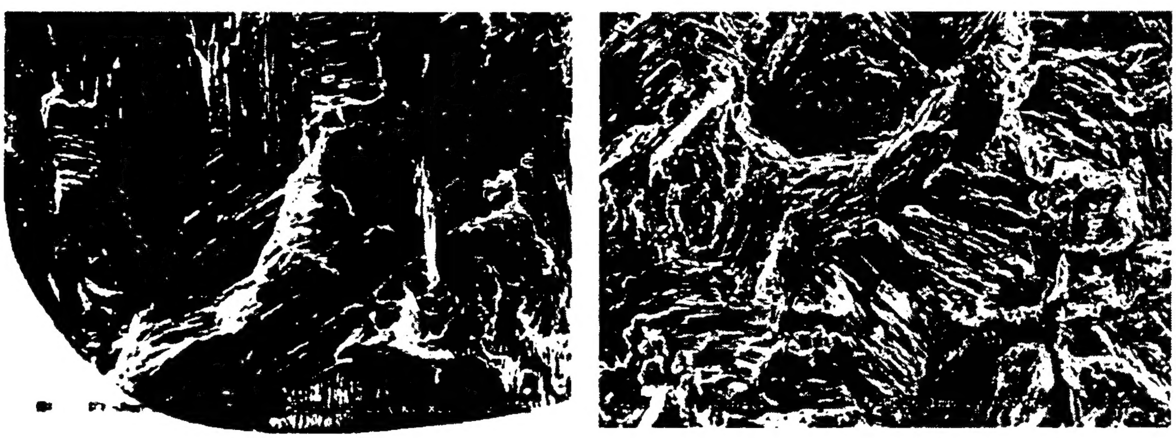
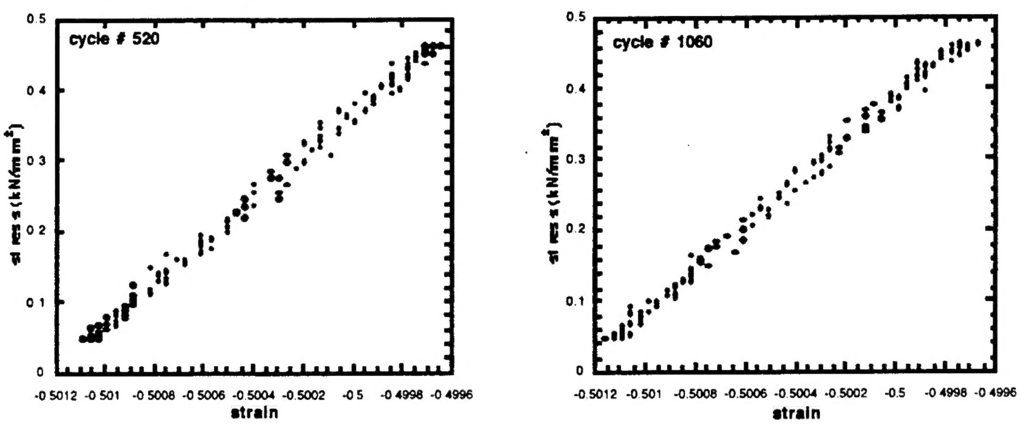


Fig. 3: Image of surface replica recorded after 5049 cycles from the specimen having a failure life (N_f) of 8274 cycles. The arrow indicates the stress direction



(a) (b)

Fig. 4: Fatigue fracture surface of the specimen tested at σ_{max} = 450 MPa; (a) crack nucleation region and (b) away from crack origin



(a) (b)

Fig. 5: Hysteresis recorded from S-N fatigue test of the specimen cycled at σ_{max} = 460 MPa; (a) at N = 520 cycles and (b) at N = 1060 cycles

IV.2 The Role of Micro-cracking on Ductility in Lamellar TiAl polycrystals

Although lamellar PST crystals show ductility ranging from 5% (hardmodes) to about 20% (softmodes) [1,2] polycrystals show only about 1-2% ductility [3,4]. In analogy with fiber composites, it can be suspected that the polycrystal ductility will be controlled by the orientation that has the lowest ductility (0° or 90° orientation). Even then, the 1-2% ductility levels are hard to explain. On the other hand, if we consider that the hardmode orientations are weakened by micro-cracks formed in other orientations (cracking caused by their relatively lower strength levels), then the low ductility is explainable on the basis of maximumstrain obtainable in a cracked solid.

The limit ductility imposed by micro-cracking can be estimated using simple fracture mechanics concepts. For simplicity, let us assume that softmode orientations yield and after little plastic deformation, crack along the α_2/γ interfaces. This could occur even in the elastic region well below the macroscopic yield. Further, let us assume that final fracture is

determined by those colonies oriented 0° or 90° to the loading axis (since these are the strongest of all the orientations, Fig. 5(a)). It then follows that the ultimate fracture strength and fracture strain should be controlled by the area fraction of 0° or 90° lamellar grains and their ultimate strength and fracture strain. Due to the presence of micro-cracks in softmode lamellar grains, the stress and strain at fracture will have to be smaller, compared to the individual behavior of 0° and 90° oriented crystals. The maximumstrain attainable before the fracture of 0° or 90° oriented grains can be estimated using known fracture mechanics equations and assuming ideal distribution of lamellar grains. This is illustrated in Fig. 5(a) with micro-cracks of projected half-length 'd' (d=lamellar colony size) are arrested on either side of 0° or 90° lamellaeof size 'd'. It is sufficient to estimate the strain in basic "unit cell" (Fig.5(b)), as a measure of ductility.

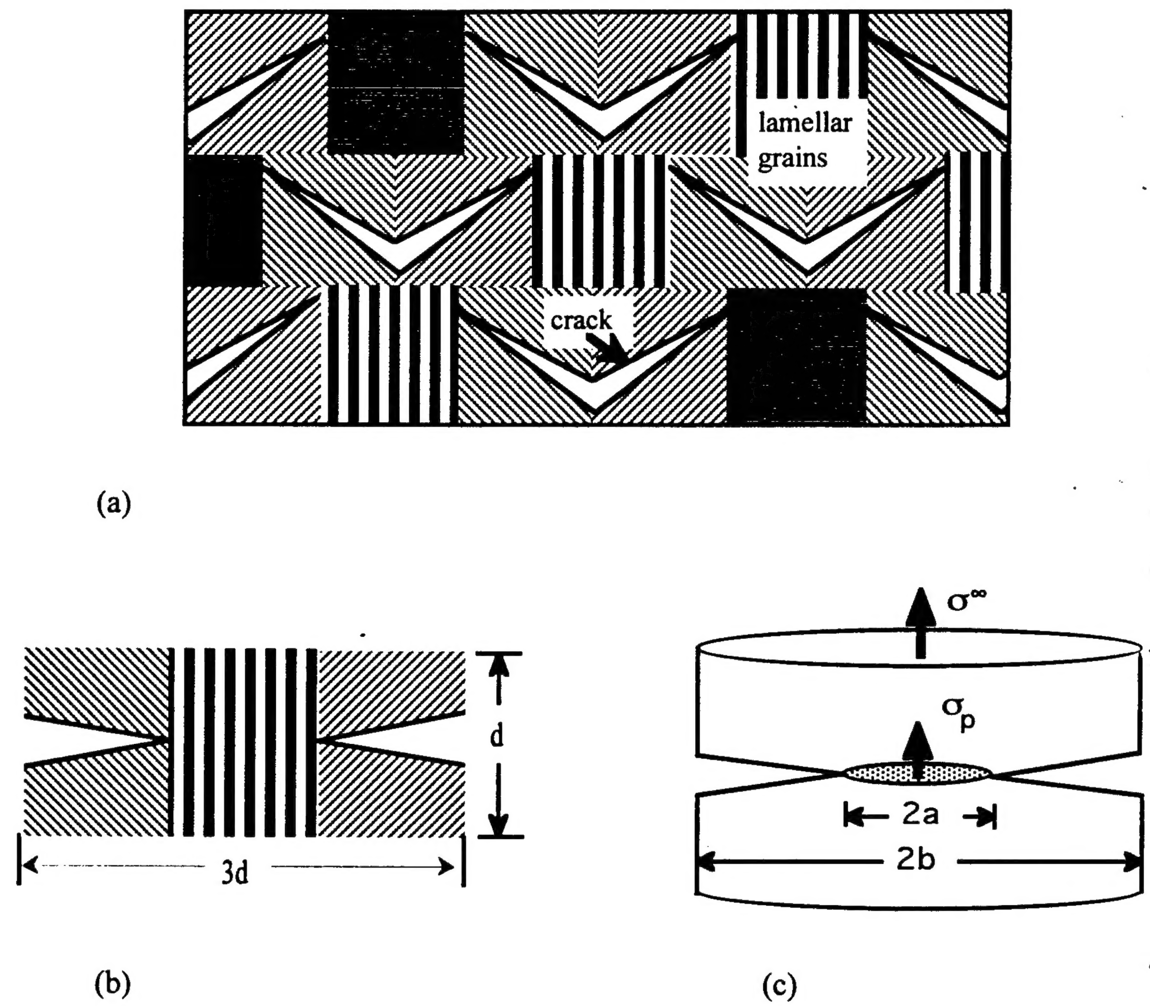


Fig. 5. Schematics of micro-crackingin lamellar Ti_3Al-TiAl polycrystals: (a) idealized crack distribution, (b) a specimen depicting a cracked ligamentand (c) the corresponding unit cell

The unit cell can be approximated to the geometry of a circular ligament of diameter '2a' in an externally cracked cylinder of '2b' outer diameter as shown in Fig.5(c). Fig. 5(a) then represents micro-cracking with area fraction of ligaments, c=0.11. The area fraction and lamellar grain size are related to 'a' and 'b' in Fig. 5(b) as: $c = (a^2/b^2)$; $a = d/2$; and $b = (d/2\sqrt{c})$. The ligament lamellae(0° or 90°) will fracture when the stress intensity factor at the crack tip reaches the fracture toughness, K_{Ic} of the 0° or 90° lamellae. The far field stress to cause this ligament fracture can be given as [5]

$$\sigma^{\infty} = \frac{2\sqrt{2}\, K_{IC}\, c}{\phi_c \sqrt{\pi d\left(1-\sqrt{c}\right)}} \qquad ...(1)$$

where $$\phi_c = \left[1+0.5\sqrt{c}+0.375c-0.363(c)^{3/2}+0.731(c)^2\right]$$

With $\sigma_p = \sigma^{\infty}c$, the average maximum displacement in the unit cell before fracture across the 0° or 90° oriented lamellae can be given by [6]

$$2u = \frac{\left(1-\vartheta^2\right)\sqrt{2\pi d\left(1-\sqrt{c}\right)}\,(1-c)\,K_{IC}}{E\,\phi_c} \qquad ...(2)$$

The fracture strain in the unit cell corresponding to 0° or 90° lamellar fracture is then given by

$$\varepsilon_f = \frac{2u}{d} = \frac{\left(1-\vartheta^2\right)\sqrt{2\pi\left(1-\sqrt{c}\right)}\,(1-c)\,K_{IC}}{\sqrt{d}\,E\,\phi_c} \qquad ...(3)$$

The total strain can be expressed as

$$\varepsilon_t = \varepsilon_f + \frac{\sigma_{prop}}{E} \qquad ...(4)$$

in which σ_{prop} is the stress at proportional limit. Since the entire micro-cracked specimen can be considered to be made up of the unit cell in Fig.5(b), the fracture strain or ductility of the entire specimen, as a first approximation, can be considered to be equivalent to the fracture strain of the unit cell as calculated from the Eqn. (4). It is to be noted that the fracture strain is inversely proportional to the square root of grain size, consistent with experimental observations which show an inverse dependence of ductility on grain size. The fracture strain from Eqn. (4) is plotted as a function of colony size in Fig. 6 with K_{Ic}=20 MPa$\sqrt{m}$; σ_{prop}=400 MPa; E=170 GPa for area fractions of 0.1, 0.25 and 0.5. It can also be seen that for a given colony size, the ductility decreases as the area fraction of the ligaments increase.

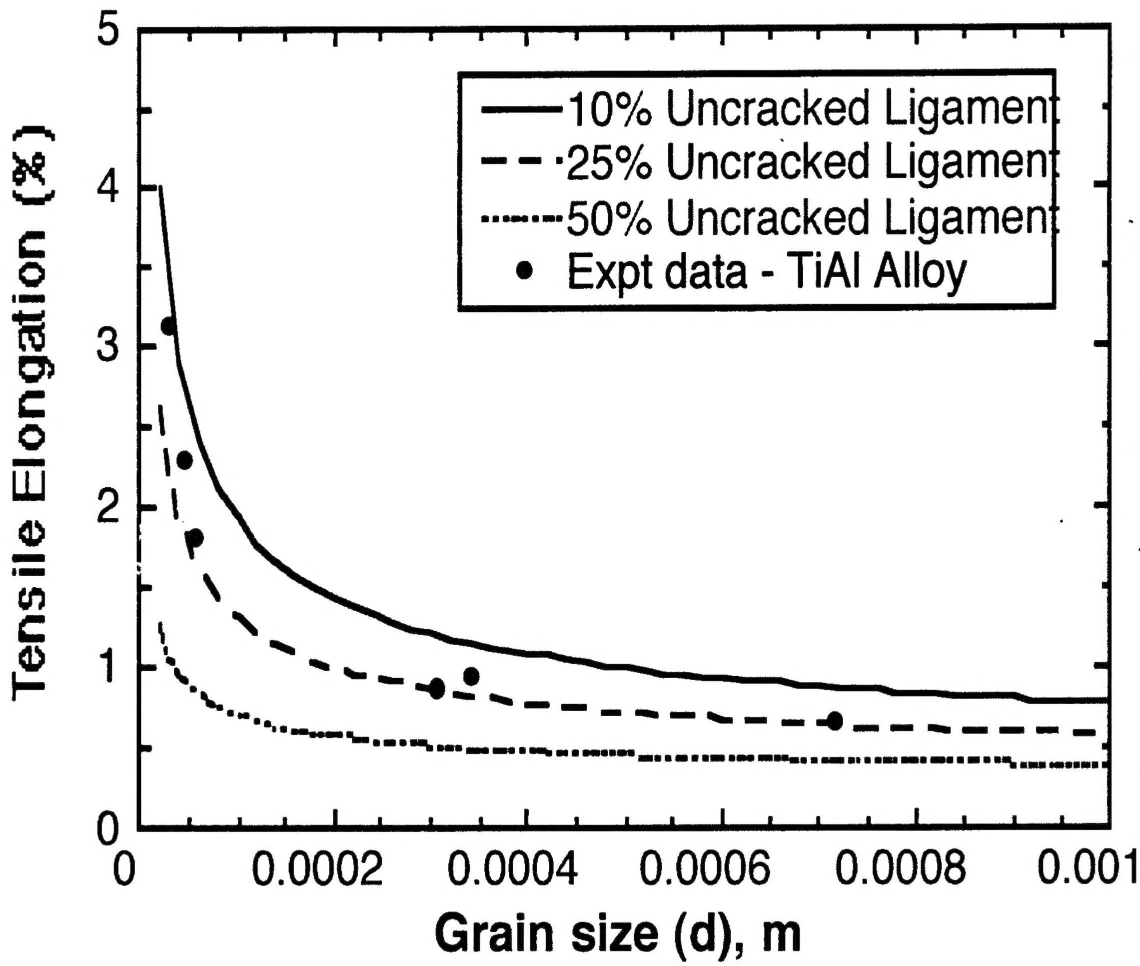


Fig. 6 Comparison of the experimental ductility data [4] with the theoretically calculated, on the basis of micro-crack limited ductility at fracture

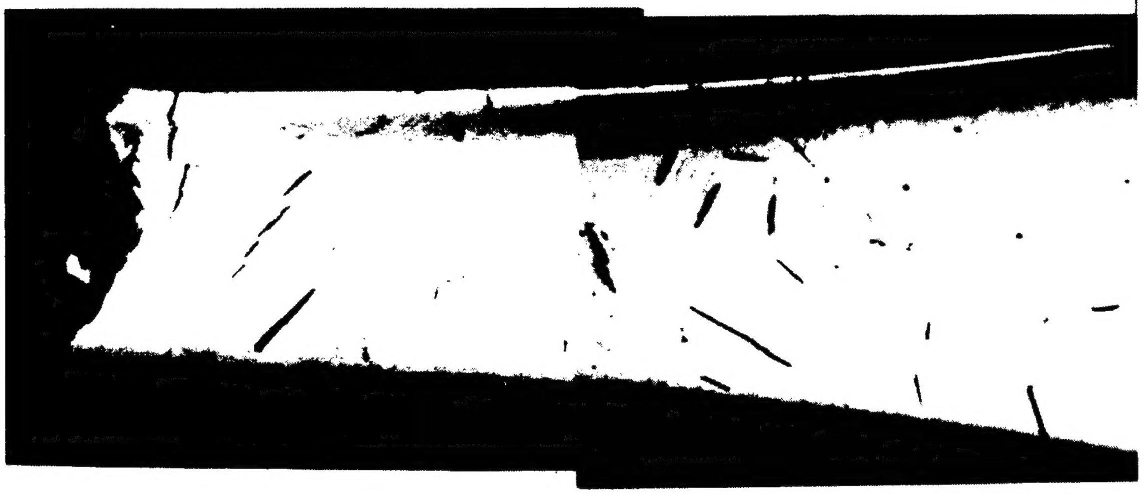


Fig. 7 Micro-cracks in a lamellar polycrystal that failed in fatigue (σ_{max} = 270 MPa, R=0.1) (100X)

From the above analysis, it is clear that micro-cracks at the lamellar grain level can limit the maximum ductility attainable in a lamellar Ti_3Al-TiAl polycrystal. This analysis is also relevant to cyclic damage of lamellar polycrystals, in which, micro-cracking was observed in the stress-controlled fatigue regime (Fig. 7).

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PUBLICATIONS

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PRESENTATIONS

None